Original Article

Comparison of residual strength behavior after indentation, scratching and grinding of zirconia-based ceramics for medical-technical applications

B. Denkena, A. Wippermann, S. Busemann, M. Kuntz, L. Gottwik

Abstract

In this study, three methods of characterizing the damage tolerance of different zirconia-based ceramics for medical-technical applications are presented. The damage is inflicted statically, with Vickers hardness impressions and dynamically by scratching with a Rockwell diamond, as well as by means of a reproducible grinding process. The damage intensity is, in each case, successively increased.

The measured strength values as a function of the inflicted damage thus provide information on the grinding robustness of the material. This permits the determination of critical grinding parameters above which the component quality is impaired and, ultimately, the patient is endangered. The continuing pressure to reduce production costs by shortening processing times makes damage tolerant behavior of materials extremely important. Ultimately, this permits the reduction of production costs while maintaining component quality and the guarantee of future patient safety.

1. Introduction

With an average from 25 to 50% of total manufacturing costs, grinding with diamond grains is the most cost-intensive process step in the production of ceramic components [1]. Depending on the grinding parameters the ceramic component can be influenced by introduced residual stresses or cracks in the surface and subsurface area [2]. Due to the demand of cost reduction, higher material removal rates during grinding increase the risk of adversely affecting the functional reliability of the ceramic component in the subsequent application. For this reason, it is important to define the critical grinding parameters for each type of material. Furthermore, it is necessary to develop ceramic materials that are more damage-tolerant during grinding.

The fundamental damage mechanisms in a ceramic material during grinding can be simplified by a static indentation test. The damage produced by a sharp or blunt diamond tip has already been investigated and explained by means of fracture mechanical models [3–6]. In order to add a dynamic component during indentation of the diamond, scratches are made and characterized on the material surface [7–10]. Lee et al. made scratches on the surface of different stabilized zirconias and characterized the inflicted damage and the influence on the residual strength behavior.

The damage mechanisms during grinding are more complex and can only be described with statistical models. At any time during the grinding process, an unknown number of diamonds with varying sharpness and at various depths is involved. Different crack systems and internal stress profiles are thus overlaid in the subsurface area, which are additionally influenced by the generated process heat [11,12]. If removed chips repeatedly get in front of the diamonds, they can fuse with the ground surface due to high pressure and temperature comparable to a sinter process (crushing) [9,13]. In addition to the effects described above, there is a phase transformation of the tetragonal zirconia during grinding of zirconia-based ceramics, which counteracts the crack initiation and propagation through additional compressive stresses arising in the subsurface area [14].

The state of damage after grinding is very often evaluated on the basis of the resulting surface quality (SEM picture and roughness measurement). However, this gives no information on the residual strength and lifespan of the ceramic components. Pfeiffer and Hollstein showed that values for the roughness produced by the grinding process do not correlate with the residual strength, but rather with the depth of damage, which can be quantified with non-destructive internal stress measurements [2]. Also, the direction of grinding and subsequent orientation in the stress field has been shown to influence the residual strength result [8,15–17].

A special feature of zirconia-based ceramics is the transformation toughening effect which was published as early as 1975 in Nature by Garvie et al. [18]. High stress peaks in the area of crack fronts lead to
locally restricted transformation from the tetragonal phase stabilized at room temperature to the monoclinic zirconia phase. Accompanied by a volumetric expansion of approx. 4%, this leads to compressive stresses in the microstructure, which stop cracking. This effect results in a pronounced fracture toughness of the material. By introducing a secondary phase, the fracture toughness can be further increased \[19-22\].

In the particular case of composite materials, internal stresses in the area of the interface between matrix components and secondary phase play a major role \[23\]. Differences in the coefficients of thermal expansion lead to tensile or compressive stresses in the interface during cooling after sintering. “Crack deflection” or “crack arrest” are effects already described by Selsing in 1961 which influence the damage mechanisms \[14,24\].

In this investigation, various zirconia-based composite materials are damaged in a defined manner and then tested for residual strength. To create the damage, defined hardness impressions and scratches are made and characterized in the material surface. In addition, the damage induced by a reproducible grinding process is evaluated. The damage tolerance of the material is then determined by four-point-bending tests. If the residual strength after damage is higher for material 1 than for material 2, the former is considered more damage tolerant (see Fig. 1). The drop in strength as a result of increasing damage intensity is, of course, overlaid by intrinsic defects in the volume of the material \[10-12\].

The present work aims to show different investigation methods to evaluate the damage tolerance behavior and grinding robustness of zirconia-based ceramics for medical-technical applications. The results are compared with one another and discussed.

2. Material and method

2.1. Material

The investigations of the damage tolerance is carried out on three high-strength, zirconia-based composite materials (Fig. 2). The zirconia-toughened alumina (ZTA) from CeramTec GmbH (Germany) is a biocompatible composite material, mainly used for high wear-resistance applications. It consists of 80 vol.% of alumina and 17 vol.% of finely dispersed, yttrium-stabilized zirconia. As a further, third phase, the ZTA contains 3 vol.% of strontium hexaaluminate. Also tested is a newly developed composite material, HTZ500, also from CeramTec GmbH. It contains 95 vol.% of yttrium-stabilized zirconia with 5 vol.% of finely dispersed strontium hexaaluminate. It is characterized by a significantly higher fracture toughness and bending strength. The third composite material is a classical alumina-toughened zirconia (ATZ) from Tosoh Corp. (Japan). It consists of 70 vol.% of yttrium-stabilized zirconia and 30 vol.% of finely dispersed alumina. The most important properties of the composite materials measured are summarized in Table 1.

From all materials, bending bars are prepared conventionally by biaxial pressing followed by three-step sintering (pre-sintering, hot isostatic pressing, post-HIP-treatment). The as-fired surfaces are ground to size on all sides and roughness (\(R_s < 2 \mu m\)) as specified in EN 843-1 with gentle parameters for subsequent strength measurements. The volume fraction of monoclinic phase on the ground surface of the bend bars are measured and outlined in Table 1 and in Fig. 16 as initial value. The tension side of the bending bars are chamfered and tested.
2.2. Damage by indentation

On the tension side of the ground bending bars, three hardness impressions with a Vickers diamond are made using a KB-30-SR tester from KB Prüftechnik GmbH under the same load. The exact position and orientation of the impressions is ensured by means of an applied template (Fig. 3). The impressions are oriented so that their diagonals were parallel/orthogonal to the direction of tension. Pre-trials for reproducibility of the method showed a decreasing scatter of the strength values with increasing indent load. For this reason, a differing number of test pieces are defined, depending on the load. 10 test pieces without hardness impressions are tested for each material, five test pieces in the load range from 1 to 30 N, four from 31 to 100 N, and three for loads above 100 N. All indented test pieces are altered afterwards for one week in distilled water to ensure saturation of the subcritical crack growth. All damaged test pieces thus have the same starting conditions for the further strength test. The strength test is a four-point bending test in accordance with EN 843-1. All damaged test pieces are then reconditioned in the grinding machine using an automated dressing process. During grinding, the process is cooled with a mixture of water with 5–10% of cooling lubricant. All specimens have been ground within one session and are exposed to comparable environmental conditions. After processing, the test pieces are released from the carrier plate, chamfered and tested for residual strength in a four-point bending test in accordance with EN 843-1.

2.4. Damage by grinding

The pre-ground bending bars are glued to aluminum carrier plates with an industrial lacquer. For this purpose, the carrier plates are provided with grooves and an overflow gutter for excess glue so that the bars can be set exactly plane parallel. The carrier plates are attached to a Kistler multi-component dynamometer screwed into an RFM 600 DS five-axis machine tool from Röders. For the grinding tests, a circumferential grinding wheel with synthetic resin bond (1A1, D91, C150) is used. The grinding process used is longitudinal-periphery grinding (see Fig. 5).

Five bending bars are ground for each parameter combination. Different material removal rates are tested by varying the depth of cut $a_k$ and the cutting speed $v_c$. Every five test pieces, the grinding wheel is reconditioned in the grinding machine using an automated dressing process. During grinding, the process is cooled with a mixture of water with 5–10% of cooling lubricant. All specimens have been ground within one session and are exposed to comparable environmental conditions. After processing, the test pieces are released from the carrier plate, chamfered and tested for residual strength in a four-point bending test in accordance with EN 843-1.

The residual strength values are plotted against the process parameters $a_k$ and $v_c$ and against the grinding power. The contact area related grinding power $P^*_{c}$ in equation 1 is calculated from the measured tangential force $F_t$, the cutting speed $v_c$ and the grinding contact area, $20$ and $25$ N. The length of the scratches is 2.5 mm and the feed velocity is 5 mm/min. Three test pieces are damaged for each load. As before, the test pieces are altered in distilled water for 7 days before the fracture stress is determined in a four point bending test in accordance with EN 843-1.

### Table 1

<table>
<thead>
<tr>
<th>Material</th>
<th>Testing method according to</th>
<th>HTZ500</th>
<th>ZTA</th>
<th>ATZ (TZ-3Y20AB)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Manufacturer</td>
<td>--</td>
<td>CeramTec</td>
<td>CeramTec</td>
<td>Tosoh</td>
</tr>
<tr>
<td>Matrix component [vol.%]</td>
<td>--</td>
<td>95% ZrO2:Y</td>
<td>80% Al2O3</td>
<td>70% ZrO2:Y</td>
</tr>
<tr>
<td>Second phase [vol.%]</td>
<td>--</td>
<td>5% SrAl12O19</td>
<td>17% ZrO2:Y</td>
<td>30% Al2O3</td>
</tr>
<tr>
<td>Third phase [vol.%]</td>
<td>--</td>
<td>2% Y2O3</td>
<td>1% Y2O3</td>
<td>3% Y2O3</td>
</tr>
<tr>
<td>Stabilizer in zirconia [mol.%]</td>
<td>--</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Content of monoclinic phase on polished surface [vol.%]</td>
<td>ASTM F 1873–98 M</td>
<td>≤1</td>
<td>4</td>
<td>≤1</td>
</tr>
<tr>
<td>Content of monoclinic phase on ground surface [vol.%]</td>
<td>ASTM F 1873–98 M</td>
<td>16</td>
<td>18</td>
<td>5</td>
</tr>
<tr>
<td>Grain size of matrix component [μm]</td>
<td>DIN EN 623-3: 2003</td>
<td>0.27</td>
<td>0.51</td>
<td>0.28</td>
</tr>
<tr>
<td>Grain size of second phase [μm]</td>
<td>DIN EN 623-3: 2003</td>
<td>0.21</td>
<td>0.28</td>
<td>0.33</td>
</tr>
<tr>
<td>Density [g/cm³]</td>
<td>DIN EN 623-2: 1993</td>
<td>5.96</td>
<td>4.37</td>
<td>5.46</td>
</tr>
<tr>
<td>Hardness HV1 [GPa]</td>
<td>DIN EN 843-4:2005</td>
<td>12.6</td>
<td>17.8</td>
<td>14.5</td>
</tr>
<tr>
<td>Indentation fracture toughness K1cIF [MPa ∙√m]</td>
<td>ISO 14627:2012-07</td>
<td>8.6</td>
<td>5.9</td>
<td>5.5</td>
</tr>
<tr>
<td>Fracture toughness K1cSEVNB [MPa ∙√m]</td>
<td>DIN CEN/TS 14425-S</td>
<td>6.9</td>
<td>5.5</td>
<td>6.5</td>
</tr>
<tr>
<td>Characteristic strength σ0.2 [MPa]</td>
<td>DIN EN 843-1:2006</td>
<td>1628</td>
<td>1417</td>
<td>1497</td>
</tr>
<tr>
<td>Weibull modulus m</td>
<td>DIN EN 843-5: 2006</td>
<td>10.1</td>
<td>26.3</td>
<td>16.4</td>
</tr>
<tr>
<td>Coefficient of expansion α [10⁻⁶ K⁻¹]</td>
<td>DIN EN 821-1:1995</td>
<td>10.6</td>
<td>8.1</td>
<td>9.5</td>
</tr>
<tr>
<td>Thermal conductivity λ [W/m ∙ K]</td>
<td>DIN EN 821-2:1997</td>
<td>3.6</td>
<td>17.0</td>
<td>6.0</td>
</tr>
</tbody>
</table>
which is given by the width of cut $a_p$ and the geometric contact length $l_g$ [25].

$$P'_c = \frac{(F_t \cdot v_c)}{(a_p \cdot l_g)} \ [W/mm^2]$$

Equation 1: Calculation of grinding power

On selected test pieces, a fractographic examination to determine the fracture origin was carried out using an S-4700 II scanning electron microscope with field emission (cold) cathode from Hitachi. A phase determination on selected, machined surfaces was carried out using a Bruker AXS D8 Advance X-ray diffractometer with Cu $K_{\alpha}$ radiation ($\lambda = 1.5406 \ \AA$). The content of monoclinic phase at the ground surface was determined using the analysis method described in the ISO 13356:2016-02 standard.

3. Results

3.1. Damage by indentation

Fig. 6 shows the residual strength values depending on the Vickers indent load. The initial strength is plotted on the ordinal axis. This includes only volume defects and reflects the quality of material production. The indent load is shown on the abscissa with a logarithmic scale.

There is a specific curve for each of the materials tested, which is characterized by a plateau region and a region of decreasing strength. In the plateau region, the volume defects dominate over the surface defects inflicted and thus determine the strength. Above a critical indent load, the surface defects become the origin of failure. The reduction of residual strength with increasing indent load is a measure of the damage tolerant behavior of a material. Here the HTZ500 material shows residual strength values that are higher than those for ZTA and ATZ by a factor of 3.5. The residual strength of the HTZ500 material after damaging with HV50 is 534 MPa, whereas, for the ZTA and ATZ materials it is 148 and 160 MPa. By comparing the $K_{IC}$ in Table 1, the values of the HTZ500 material are only higher than those for ZTA and ATZ by a factor of approximately 1.5. A possible explanation are different residual stress states in the subsurface area which are controlling the crack growing, f. e. stress-free areas after polishing or overall compressive stress due to grinding.

A comprehensive fractographic examination for the fracture origin was carried out. Taking the ATZ material as an example, Fig. 7 shows the fracture origin after inflicted damage with HV3, HV30 and HV50. Fractograph A shows a typical volume defect located in the strength plateau region, not affected by indentation. The volume defect is a porous region without chemical abnormalities and has a width of approximately 15 $\mu$m. Fractographs B and C show the hardness impressions on the surface to be fracture origins. In fractograph C, a median crack can be seen to extend below the impression, reaching about 500 $\mu$m into the volume of the material. The residual strength values found correlate with the fractographic photos.

3.2. Damage by scratches

Fig. 8 shows the residual strengths as a function of the applied normal force when scratching the ceramic surface with the Rockwell diamond. The curves for the three materials differ greatly. As for damage by indentation, the HTZ500 material shows a plateau region and a region of decreasing strength. In comparison with ZTA and ATZ, it shows the highest initial strength. The scratches made with a normal force of 10 N do not reduce the strength of HTZ500. Increasing the normal force from 15 to 25 N reduces the strength from 1542 to 1081 MPa. The behavior of ATZ is similar, apart from a decrease in strength from its initial value of 1447 MPa to 1058 MPa with a normal force of 10 N. Both materials meet the strength requirements of ISO 13356 for implant materials based on yttrium-stabilized zirconia.

The material ZTA shows a significantly greater decrease in strength as a result of the scratches. With a normal force of 10 N, the strength has already fallen from its initial value of 1386 MPa to 500 MPa. Further increasing the normal force to 15 N reduces the strength to 161 MPa. Damage with even higher forces up to 25 N only reduces the strength slightly to a minimum of 133 MPa. All inflicted damage on the ZTA by scratching led to values below the ISO 13356 requirement of 800 MPa.

A fractographic examination for the fracture origin was carried out. All scratches were identified as fracture origin. No difference in fracture surface between the materials could be identified.

Fig. 9 shows optical micrographs of inflicted scratches on polished surfaces with a normal force of 25 N. The micrographs were taken of polished surfaces since this visualizes the damage more clearly. The HTZ500 and ATZ materials show only slight damage at the edge of the scratch. The damage intensities are more or less comparable. ZTA, on the other hand, shows a marked tendency to chipping along the scratch. These results correlate with the residual strength curves in Fig. 8.

Fig. 10 shows SEM micrographs of the scratches inflicted with a normal force of 25 N. The materials HTZ500 and ATZ show only very slight damage along the scratch. Plastically deformed areas are clearly visible at the bottom and on the flanks of the scratch. In addition to plastically deformed areas, the ZTA micrograph shows wide ranging damage in the form of cracks and chipping.

3.3. Damage by grinding

For the purposely damage of the surface and the subsurface area,
bending bars were ground on the side that come subsequently under tension. For the first series of tests, the depth of cut was successively increased from 0.02 to 0.18 mm during grinding. The other process parameters were not changed.

In Fig. 11, the measured residual strengths are shown as a function of the depth of cut during grinding. The initial strengths of the various materials are plotted on the ordinal axis. Comparing the strength curves of the materials, two trends can be identified. The strengths of the HTZ500 and ATZ materials fall significantly with increasing depth of cut during grinding. The ZTA material shows an opposite trend. At low values, the strength of ZTA shows a significant drop to 1100 MPa but then rises with further increase of depth of cut to return close to its initial value. The residual strength curve of ZTA thus intersects the curves for HTZ500 and ATZ. This fundamentally different material behavior when damage is inflicted by grinding does not reflect the results of the indentation and scratch tests presented previously.

Fig. 12 shows the calculated normal forces during grinding for test series 1. There is no anomaly to be seen that permits conclusions to be drawn regarding the strength result in Fig. 11. With increasing depth of

---

**Fig. 7.** Results of fracture origin analysis, with ATZ as example, after damaging with HV3 (A), HV30 (B) and HV50 (C).

**Fig. 8.** Residual strength of zirconia-based composites as a function of the scratch load (dashed line marks the required standard of 800 MPa according to ISO 13356).

**Fig. 9.** Scratch overview and damage intensity on surface of zirconia-based composites (25 N scratch load, 5 mm/min feed velocity).
cut during grinding, the normal force during grinding increases according to the material hardness HV1. The harder the material, the greater is the resistance of the material to the penetrating diamond during the cutting process.

Fig. 13 shows the residual strength results from test series 2. Here, the cutting speed was varied from 10 to 50 m/s, other process parameters remaining constant. The resulting strength curves confirm the results from test series 1. However, at low cutting speeds, plateau regions can now be seen in the cases of the HTZ500 and ATZ materials, in which the damage by grinding is smaller than the volume defects in the material. At low cutting speeds, the strength of the ZTA material is even lower than that of the HTZ500 and ATZ materials. Here, too, differences between the curves for HTZ500 and ATZ and that for the ZTA material are clearly to be seen. This effect has already been described in test series 1 and can now be considered to be significant. Overall, the strength results for the HTZ500 material from test series 1 and 2 are well above the permissible minimum for implant applications according to ISO 13356.

A fractographic examination was carried out at cutting speeds of 20 and 50 m/s (see marked areas A and B in Fig. 13). At both cutting speeds, the HTZ500 and ATZ materials showed surface damage as fracture origins as a consequence of the grinding process. In the case of the ATZ material, the fractograph at a cutting speed of 20 m/s indicates a fracture origin deep within the volume of the material. The actual surface defect could not be found. This is a secondary fracture. Both of the ZTA test pieces indicate large area cracking in the region of the chamfer as fracture origin. During the grinding tests, chipping was already noted at the edges of the test pieces. Deep-reaching edge cracks could not be completely removed by the subsequent chamfering of the test pieces. Overall, damage as a consequence of the grinding process could be demonstrated on all fracture surfaces (Fig. 14).

Fig. 15 shows the measured normal forces as a function of cutting speed. There is a trend opposite to that of test series 1 (variation of depth of cut). The normal forces during grinding decrease exponentially with increasing cutting speed. Increasing the cutting speed increases the number of diamond grains coming into contact with the material per unit time. More diamond grains participate in the cutting process per time unit. The grinding forces are reduced and as a consequence the
To this end, three different kinds of damages were inflicted on the material surfaces. In the first case, hardness impressions with a Vickers diamond were made statically with various loads; in the second case, dynamic scratches were made with a Rockwell diamond and in the third case, damage was inflicted by means of a reproducible grinding process. Only limited correlation between the resulting residual strength curves was found to be possible. The damage by indentation and scratching is localized. The damage by the grinding process, on the other hand, is distributed over the entire tension side of the ceramic bars and exhibits overlaps. Additional internal stress profiles induced by matrix distortions or phase transformations in the area of the damage are thus either concentrated close to the defect or, in the case of the grinding process, spread over the whole subsurface area. In addition, the formation of the internal stress profiles and the generation of damage by the grinding process are influenced by temperature effects. In the grinding process, almost all the mechanical energy is converted to thermal energy. In a parallel study, temperature measurements during grinding were carried out with the same process parameters and materials. Peak temperatures well above 1000 °C were found in the process zone. These temperatures are in the reversion transformation range of monoclinic zirconia [26].

In Fig. 17 the residual strength was plotted as a function of the contact area-related grinding power to be able to discuss better the additional temperature effect on the damage by grinding. The correlation coefficients R² for the various materials are rather unsatisfactory. However, differing material behaviors can be derived from the curves. The HTZ500 and ATZ materials show decreasing residual strength with speeds of 10 and 50 m/s during grinding. To complete the picture, the initial values before grinding are also shown (see Fig. 16).
increasing grinding power, although, in comparison to ATZ, all calculated residual strength values for the HTZ500 material remain above the lower limit of 800 MPa required by ISO 13356 for implant applications. HTZ500 thus shows more robust behavior in the grinding process. This behavior is reflected in the previously presented results of the indentation and scratch tests. The ZTA material showed completely different behavior. The residual strength increases steadily with increasing grinding power — but starting from a lower level. All residual strength values meet the requirement of ISO 13356. It has thus been shown that the ZTA material is damage tolerant to grinding but not to indentation or scratching.

By comparison of the results in Fig. 17 with the monoclinic phase content found at the ground surface (Fig. 18) there is the same trend for all materials. The grinding induced phase transformations in the subsurface area are inversely proportional to the grinding power.

Since the initial values of the different materials are slightly smaller, it can reasonably be assumed that grinding induced phase transformation first increases until a critical grinding intensity is reached and then decrease drastically with any further increase in intensity. Two possible approaches to explain this are now described:

1. Complete areas of microstructure are torn out with hardly any plastic deformation or phase transformation events. The dominant removal mechanisms are chipping and crushing.

2. High process temperatures lead, in part, to a reverse transformation of the monoclinic phase to the tetragonal zirconia phase. Internal compressive stresses are thus relieved in the subsurface area.

A correlation between residual strengths and monoclinic phase content is only found in the HTZ500 and ATZ materials. Assuming that the grinding induced phase transformation and matrix distortion in the subsurface area are the main source of resulting compressive stresses, these oppose the subsequent tensile load in the bending test. For the ZTA material, this relationship does not apply. Here, we find higher residual strengths with decreasing monoclinic phase content. It has not yet been possible to establish the cause of this opposite behavior. Internal stress measurements are currently being carried out for further explanation. Furthermore, comparative measurements of high-temperature properties of the materials are essential to an explanation. Because of its thermal properties, ZTA is clearly distinct from the HTZ500 and ATZ materials. ZTA shows a lower coefficient of thermal expansion and a thermal conductivity that is higher by a factor of 5 than that of HTZ500 and by a factor of 3 than that of ATZ, because of the higher alumina content in its structure. It can therefore be assumed that the damage mechanisms at high temperatures in the ZTA material are fundamentally different from those in the HTZ500 and ATZ materials.

5. Conclusions

The residual strength curves for the newly-developed HTZ500 material from CeramTec (Germany) showed not only the highest initial strength (without damage) but also the highest strength values in comparison with the ATZ and ZTA materials after damage by indentation and scratching. It could be identified as the most damage tolerant material. The residual strength curves following damage by a reproducible grinding process did not yield results that agreed with the indentation and scratching tests. Here, it was found that the ZTA material did not exhibit, as in the indentation and scratching test previously carried out, the lowest strength. It even showed itself extremely tolerant to damage by grinding, even achieving rising strength values with increasing grinding intensity. An examination of monoclinic phase content on ground surfaces was carried out to trace back the measured strength values to grinding-induced compressive stresses in the subsurface area. This showed that, with increasing grinding intensity, the monoclinic phase content in the subsurface area first increased and then decreased abruptly. How the different materials behave at high temperatures when subjected to grinding remains unclear. However, key elements in finding the explanation will certainly be the determination of high-temperature properties of the materials and the measurement of internal stress profiles in the subsurface area that originate at depth.

Acknowledgments

This project was funded by the Collaborative Research Center 599 for Biomedical Technology, a Center of the German Research Foundation (DFG), within the project T5 “Development of Methods for the Production and Automated Processing of Damage Tolerant, Sintered ZrO2 Ceramics for Dental Applications”, a cooperation between the Institute of Production Engineering and Machine Tools of Leibniz Universität Hannover (IFW Hannover) and CeramTec GmbH. All ceramic specimens used in this work were provided by CeramTec GmbH.

References


